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20. ABSTRACT (Continue on reverse side if necessary and identify by block number)

The influence of microstructure, slip behavior, and gaseous environment on the fatigue crack propagation of Ti alloys has been investigated. Large grain or single crystal material has been used extensively in order to ensure that the crack-tip plasticity is smaller than the scale of the microstructure. As a result, extensive Stage I crack propagation occurs. The experimental relationships between crack-tip plasticity, orientation of the crack plane to the stress axis, fractography, and the resulting fatigue crack growth behavior

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20. Abstract (continued)

is presented. Several Ti alloys are examined in lab air, dry helium, or hydrogen environments: beta-phase Ti-40V, metastable beta Ti-27V and Ti-30V, alpha-phase Ti-4Al, and the alpha-beta alloy, Ti-8Al-1Mo-1V. The behavior of the beta Ti alloys can be analyzed primarily in terms of crack advance by an intersecting slip process resulting from flow on at least two active slip systems. In contrast, the mechanism for Stage I fatigue in alpha and alpha + beta Ti alloys appears to be dominated by basal slip and strongly influenced by environment.

Aspects of the cyclic stress-strain behavior of beta and alpha + beta Ti alloys are also reported. In addition, strengthening of Ti alloys by eutectoid decomposition is briefly discussed.

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Fatigue Crack Propagation in Ti alloys

INTRODUCTION

The resistance to fatigue failure is an important design criteria of many Ti alloys. Because most Ti alloy components contain pre-existing flaws, understanding the fatigue crack propagation (fcp) behavior is critical not only in predicting fail-safe behavior but also as a basis for developing alloys or processing techniques which will lead to improved fatigue resistance. Experimental evidence indicates that fatigue crack propagation in both α(hcp) and α-β(bcc) Ti alloys is sensitive to microstructure.[1-6] The degree and manner in which the microstructure influences fatigue crack propagation appears to depend on the extent of the anisotropy of the flow behavior at the crack tip. As a result, the effect of microstructure is usually most pronounced whenever the crack tip plastic zone is smaller than the scale of the microstructure and therefore whenever crack-tip plasticity reflects "single crystal" flow behavior. This situation occurs at small stress intensity ranges and near-threshold crack growth rates in materials with small grain sizes. Microstructure also influences crack growth behavior over a wide range of growth rates in materials with a large grain size. The latter situation occurs, for example, in β forgings or in the heat affected zone of thick section Ti alloy weldments. It should also be noted that the environmental effects on the fatigue behavior are pronounced in the small crack growth region[6,7] when crack-tip plasticity is anisotropic.

The purpose of this research is to understand the influence of microstructure, slip behavior, and (to an extent) environmental effects on fatigue crack propagation in Ti alloys. Large grain or single crystal material has been used extensively, but not exclusively, in order to ensure that the extent of the crack-tip plasticity is smaller than the scale of the microstructure. As a consequence of the anisotropic slip behavior at the crack-tip in such material, the resulting crack often propagates on a plane inclined to the stress axis. Such cracking may be denoted Stage I

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crack propagation but should be distinguished from that extension of crack

initiation (not examined here) in which a "Stage I" crack propagates within

a persistent slip band generated prior to crack formation.

DISCUSSION

With the preceding as a basis, this final report summarizes progress made

in the following aspects of the program. Some of the research is still in

progress and discussion of these particular areas is limited by the data

available at this time. Since the deformation behavior of the hcp α-phase is

distinctly different than that of the bcc β-phase, the resulting fatigue behavior

is substantially different. Thus for convenience, this discussion will first

focus on the behavior of β or near-β Ti alloys and then examine crack propagation

in α and near-α Ti alloys with the β alloy behavior as a background.

a. Fatigue Crack Propagation in β Ti Alloys

Ductile fatigue crack propagation is often modeled in terms of alternating

slip or a slipping off and blunting process on a pair of intersecting slip

planes.[8-10] According to these models, the crystallography of slip should be

an important factor in determining the fatigue crack propagation behavior. The

crystallography of slip can be controlled through the use of single crystals in

which the choice of the orientation of the specimen axis influences crack tip

plasticity such that relationships between slip and fatigue crack propagation

can be examined. The resulting behavior thus represents a conflict between what

the crack-tip stress field tries to dictate and what the crystallography of slip

permits.

The "stress-field/crystallography of slip" conflict is important in much of

this research program, and it is a central theme in the study of fatigue fracture

and slip processes in a solution solution β-phase Ti-V alloy by Carlson and

Koss.[11] In this work, oriented Ti-40V single crystals or polycrystalline

specimens exhibit both the fatigue limits and the fatigue crack propagation rates which are dependent on the crystallographic orientation of the specimens, and this is shown in Fig. 1. Extensive cracking on a plane inclined to the stress axis is observed in crystals with specimen axis near the [110] direction. Fatigue crack propagation rates (da/dN) in these samples as well as those with specimen axis near the [111] direction and in polycrystalline specimens obey a dependence on the stress intensity range (ΔK) such that $da/dN = C \Delta K^{1.8}$. As will be discussed later, the constant C depends on the slip processes at the crack-tip and therefore is also a function of specimen orientation.

An examination of the features of the fracture surfaces in terms of the slip processes at the crack-tip (depicted in Fig. 2) indicate that ductile fatigue crack growth occurs by slip on intersecting slip planes and that the crack plane contains the line of intersection of at least two active slip planes. As an example, Fig. 3 shows the fracture surface of the 110 orientation sample in which most striations lie parallel to the <110> directions, which corresponds to inter-secting {211} slip planes (see Fig. 2). Taking into account the crystallography of slip in Fig. 2, rather complex arrays of striations for different samples can be interpreted in terms of the intersecting slip concept.

If fatigue crack propagation occurs by slip on intersecting planes, then it is reasonable that the relative amount of slip on each of the slip planes controls the plane of cracking. An analysis schematically depicted in Fig. 4 shows that extensive crack growth on a plane inclined to the stress axis can be a consequence of unequal slip at the crack-tip and, at least in part, on the unequal crack-tip opening displacement of the upper and lower fracture surfaces. As a result, the mechanism of this form of Stage I cracking is similar to ductile Stage II crack propagation.

All of the above observations suggest that ductile fatigue crack propagation is controlled by crack-tip opening displacements (CTOD).[12-15] Crack tip
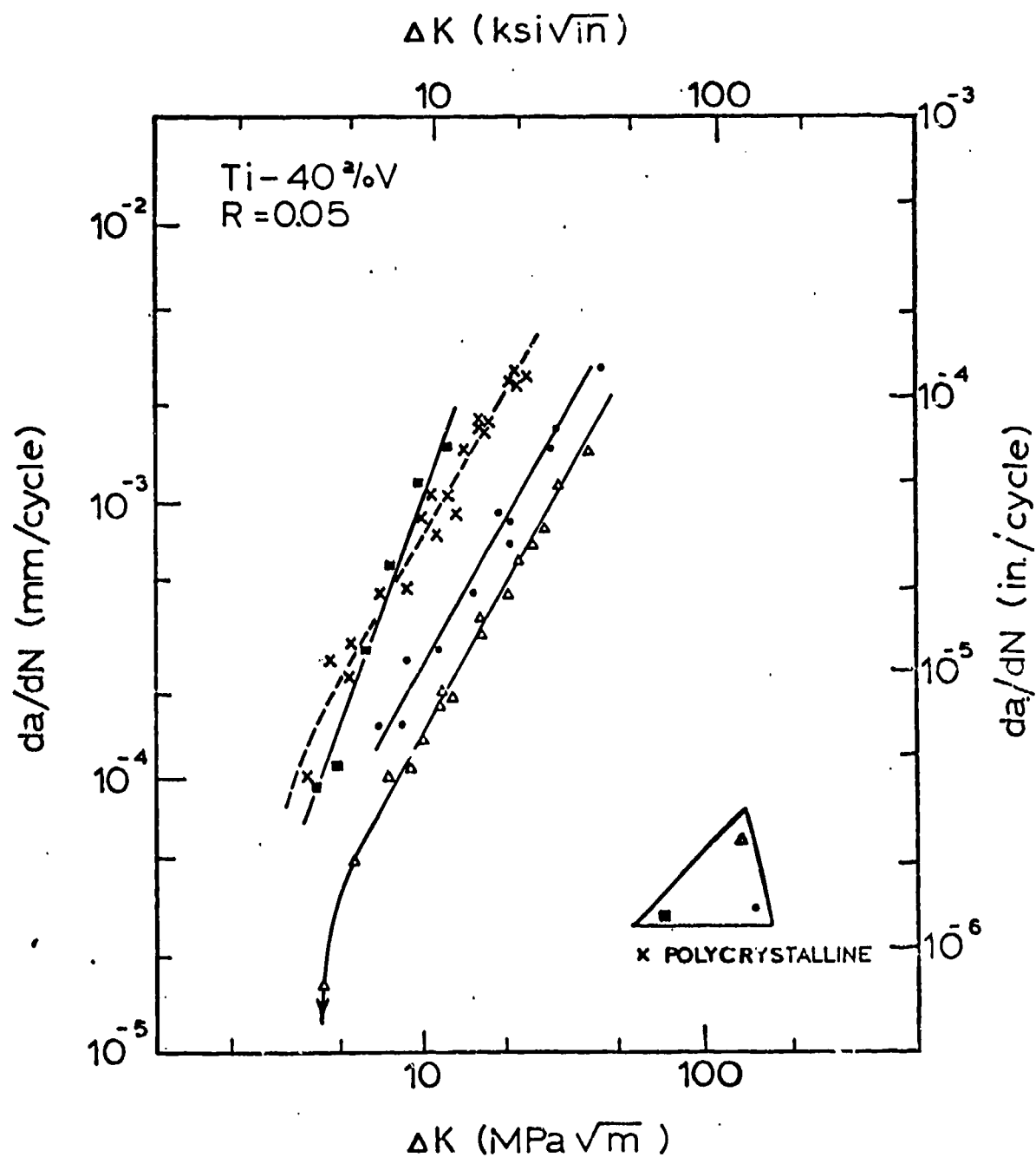
Figure 1. The fatigue crack propagation rate (da/dN) as a function of
the stress intensity range (ΔK) for Ti-40 a/o V alloy single
crystals and polycrystalline samples tested at 30 Hz and in
laboratory air (R.H. < 30%). The orientations of the stress
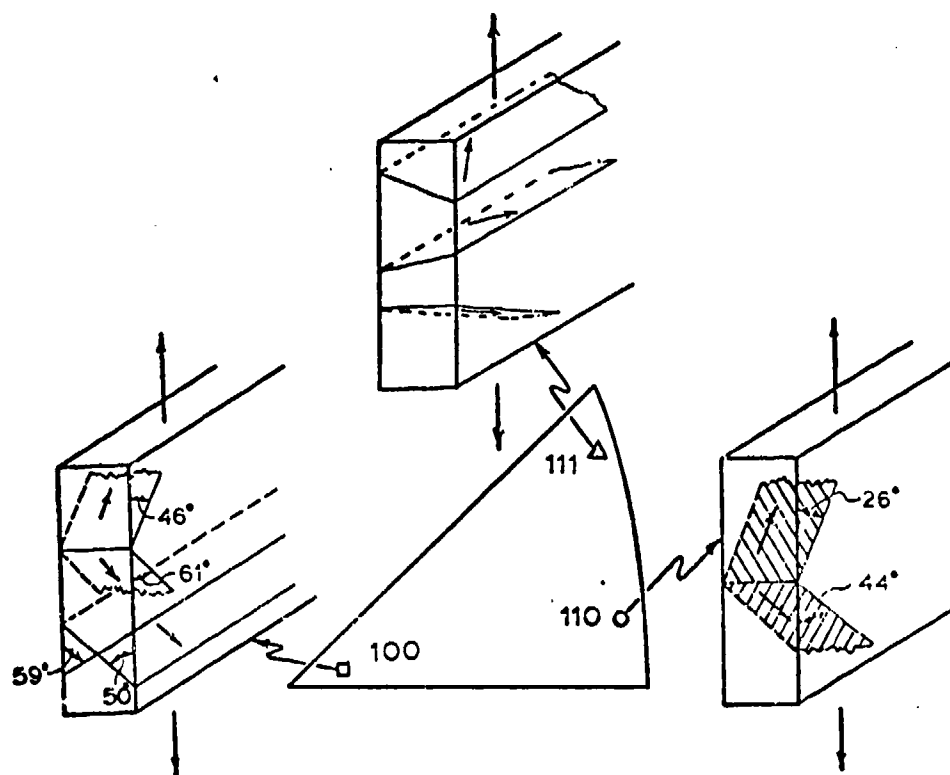axes are indicated.

Figure 2. The designations of the orientations of the single crystal specimen axes and of the most highly stressed {211} slip planes corresponding to each specimen orientation.
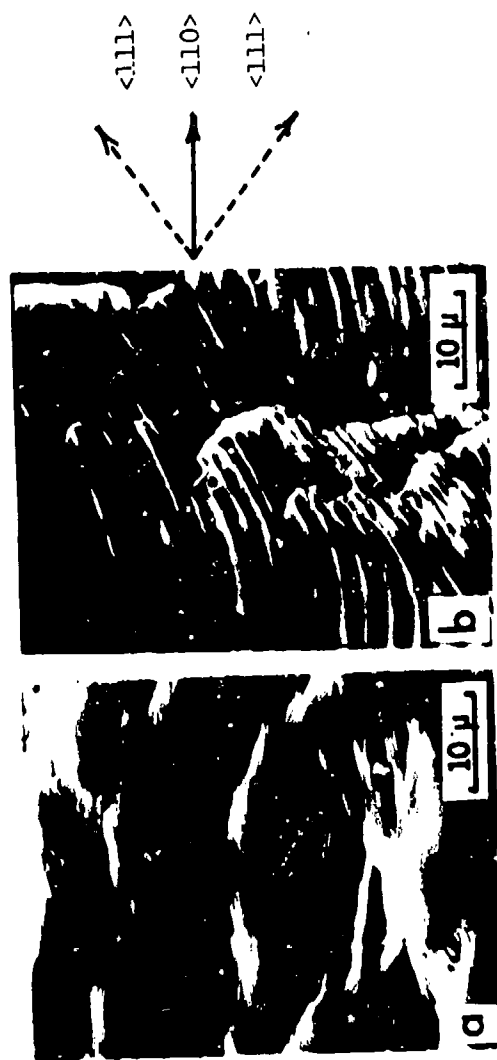
Figure 3. Electron micrographs showing the fracture surface of the 110 orientation samples at: (a) $\Delta K \approx 11$ MPa$\sqrt{m}$ and (b) $\Delta K \approx 23$ MPa$\sqrt{m}$. The crystallographic directions indicated represent lines of intersection of pairs of slip planes active at the crack tip.

plasticity and the resulting CTOD have been analyzed in terms of either (a)
assuming small-scale yielding occurs by slip on all of the planes of the two
most highly stressed slip systems ("diffuse" slip) as shown in Fig. 4a, or (b)
confining slip to sliding on two discrete bands at angle θ_1 and θ_2 to the crack
plane ("discrete" slip) as depicted in Fig. 4b. Knowing the extent of the
plastic zone, one can calculate the total opening of the crack-tip between the
upper and lower crack surfaces and relate this to the crack advance per cycle
(see ref. 11 for details). The predicted crack growth rates obey the relation-
ship:

$$da/dN = C \, \Delta K^2 \qquad \qquad (1)$$

where the magnitude of the constant C depends on whether diffuse or discrete
slip is assumed. A comparison of predicted values to those experimentally
observed for both single and polycrystalline samples shows good agreement for
all samples except the 100 samples in which fractography indicates a combination
of ductile fatigue and tensile fracture.

Two aspects of the above study on the bcc Ti-V alloy deserve special mention.
First, because of the many slip systems available in a bcc metal, isotropic
plasticity predicts behavior which is not greatly different from that described
by more rigorous single crystal plasticity described above. Secondly, most of
the results from the bcc alloy reinforce the mechanisms of ductile fatigue
cracking developed mainly on the basis of f.c.c. metal behavior. This does not
appear to be the case with α-phase(hcp) Ti alloy crystals, as will be discussed
later.

In the study of the solid solution Ti-40V alloy, slip processes were related
to fatigue crack growth behavior. It is well accepted that second phases influ-
ence both the fracture and slip behavior of the material. As a result, second
phase particles could affect fatigue fracture not only in a direct manner, such
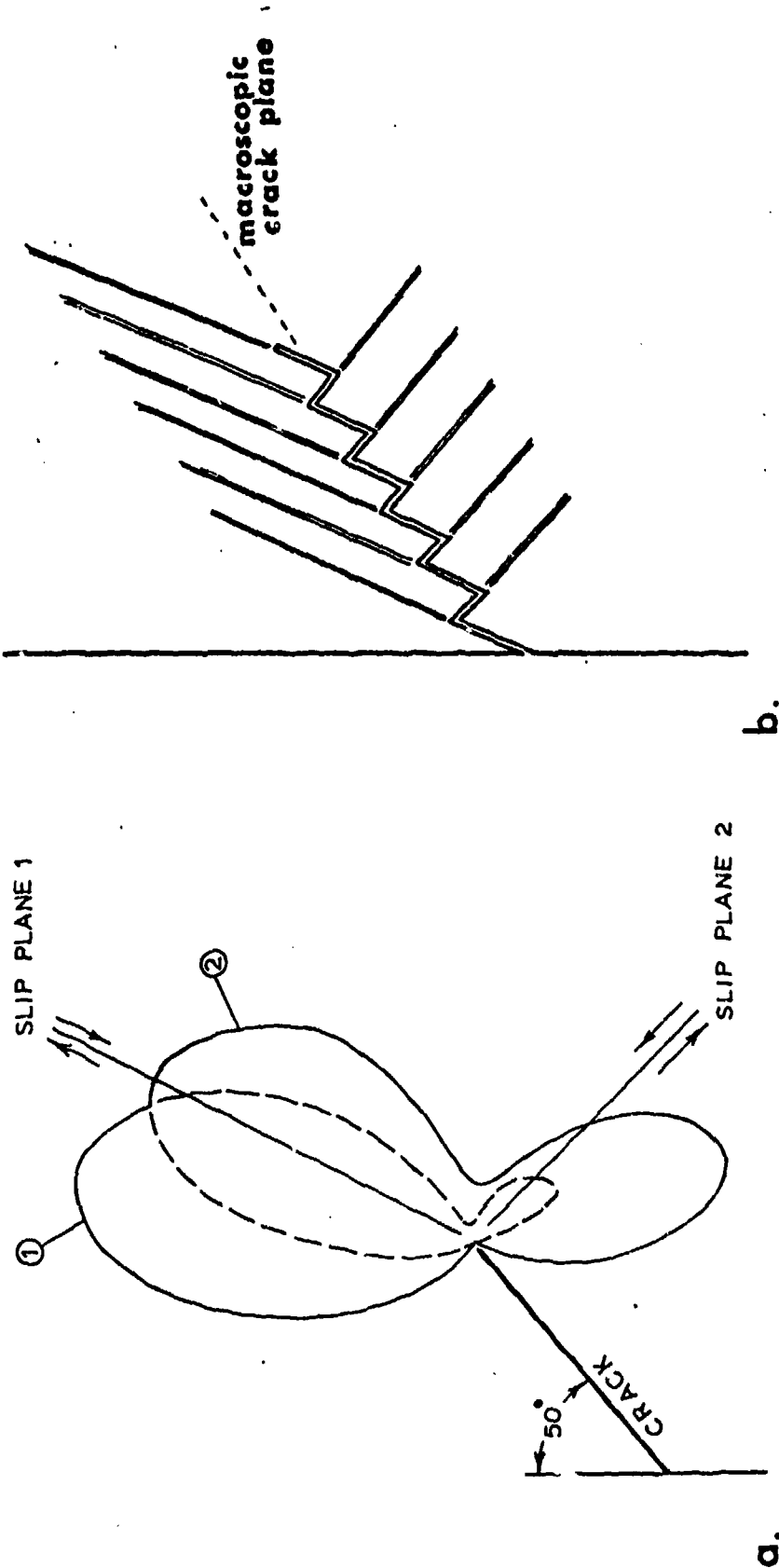as by providing a preferred crack path along the second phase, but also indirectly

Figure 4. A schematic diagram of fatigue crack propagation occurring by (a) diffuse slip or (b) coarse slip process involving unequal slip on two intersecting slip bands.

by influencing slip and therefore crack-tip plasticity. Utilizing a metastable β-phase Ti~28V alloy both as single crystal and polycrystal specimens, we have also examined the influence of microstructure of fcp behavior under conditions in which the morphology, degree of coherency, and distribution of precipitates are controlled.[16] The phase transformations in these alloys are well established, and in this study the alloys were age-hardened by heat treatments designed to precipitate either a uniform distribution of very fine, coherent ω-phase particles or alternatively large α-phase particles of different structures and morphologies. To our knowledge, this is the first published study of fatigue crack propagation in multiphase β Ti alloys. It is also the first study contrasting single and polycrystalline fatigue crack propagation behavior in any age hardenable bcc alloy, particularly under conditions where distinct difference in slip behavior can be generated by age hardening.

Oriented Ti-27V alloy single crystals and Ti-30V polycrystalline specimens have been tested after the following heat treatments: as-quenched (solid solution β), aged at 300°C (β + ω-phase), or aged at either 400° or 450°C (β + α-phase).[16] The resulting age hardening and slip behavior have been correlated with microstructure and were related to the fatigue crack propagation rates and fracture behavior. Fig. 5 shows that, except for the specimens in which plate-like α-phase appears to enhance crack growth, the single crystal samples exhibit ductile fatigue fracture with only a small influence of microstructure on the fatigue crack propagation rates. In these conditions, crack growth is adequately described by the crack-tip opening models for crack propagation previously discussed.[11-15] Unlike the single crystals, age hardening the polycrystalline samples, especially by ω-phase formation, can significantly increase the crack propagation rate, and this is shown in Fig. 6. The faster growth rates are associated directly with the presence of second phase particles, as in the case of grain boundary α-phase causing intergranular fracture after aging at
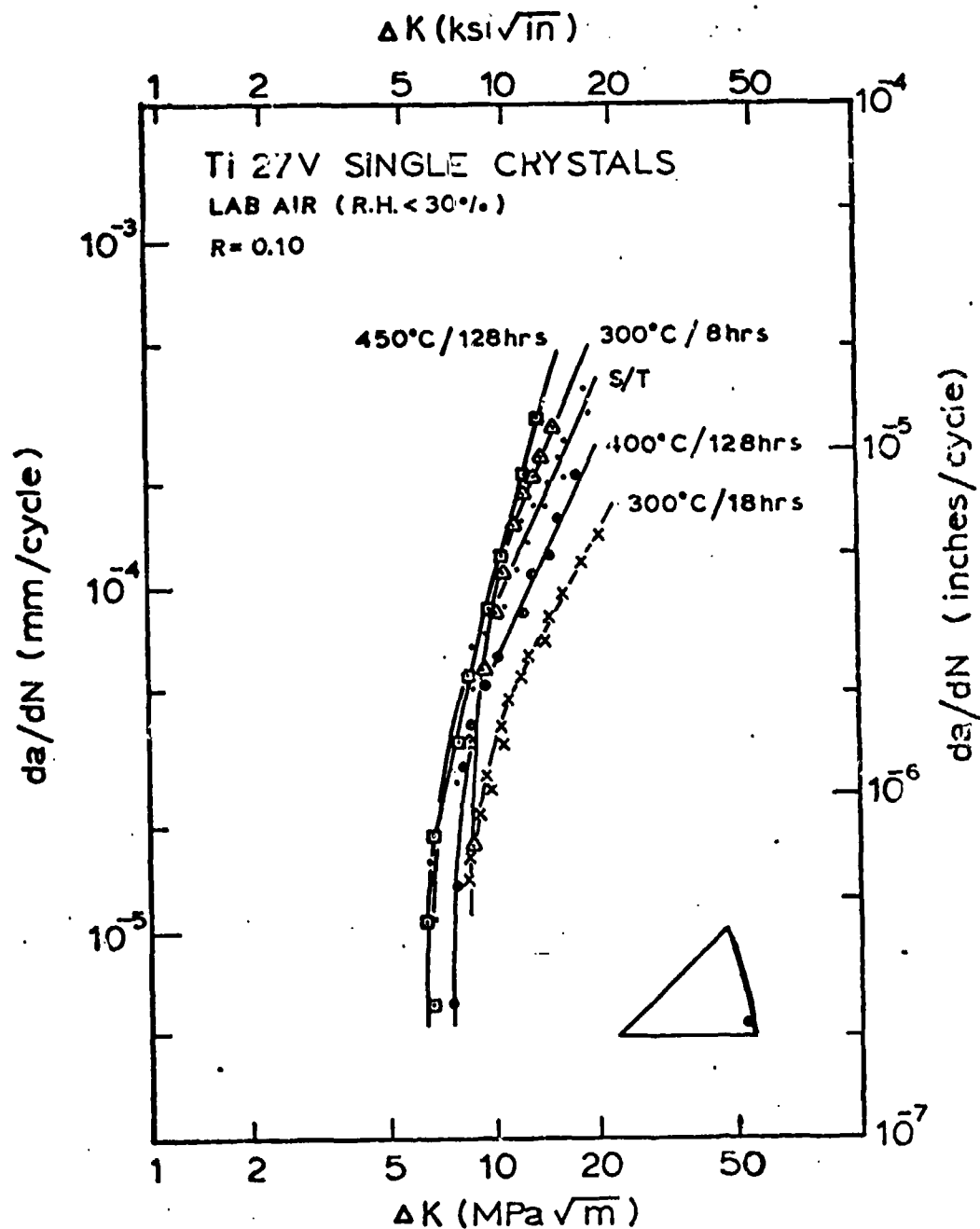
Figure 5. The fatigue crack propagation rate da/dN as a function of the stress intensity range ΔK for Ti-27 a/o **V** alloy single crystals quenched from 800°C and aged as indicated.
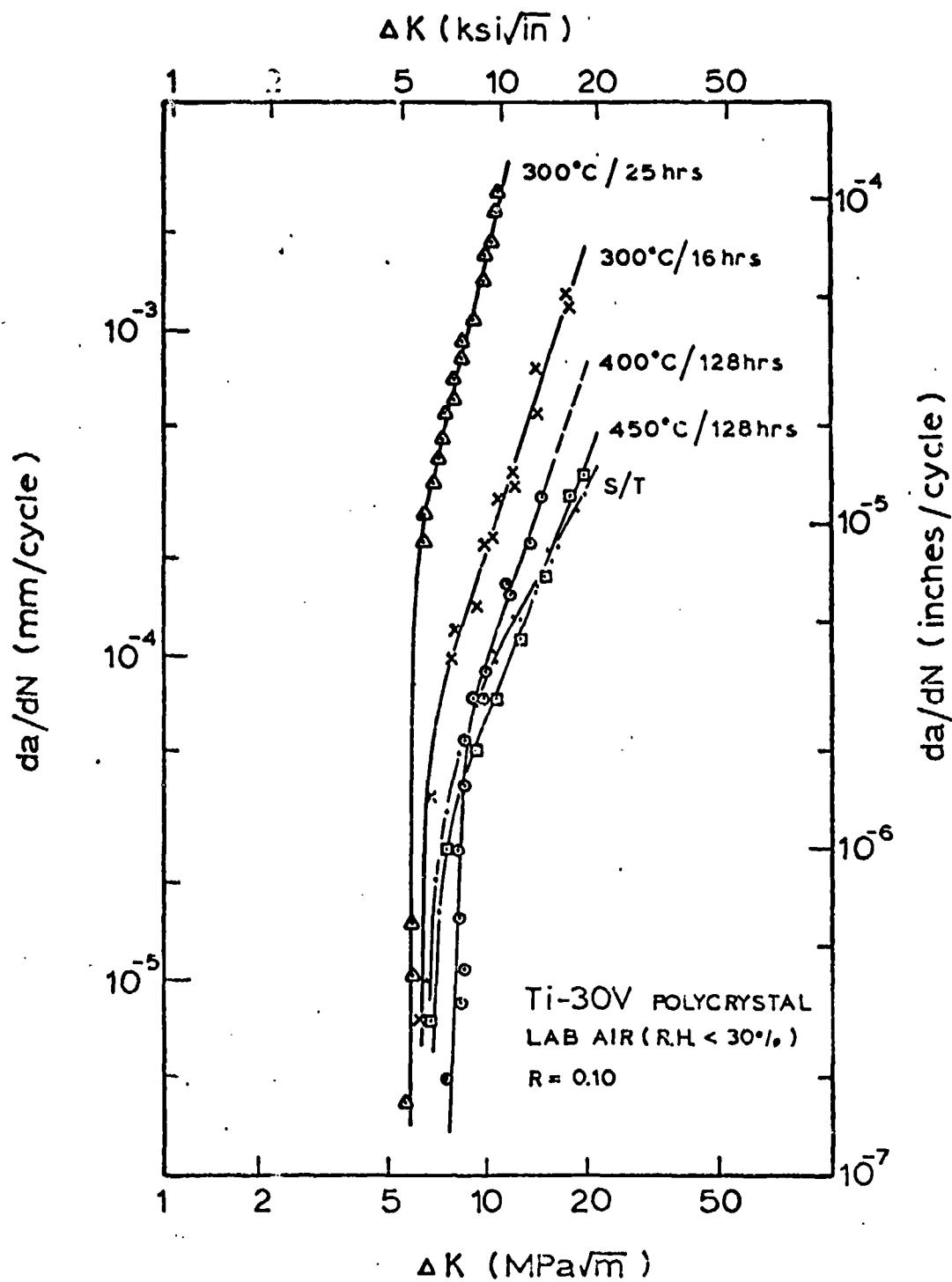
Figure 6. The fatigue crack propagation rate da/dN as a function of
the stress intensity range ΔK for Ti-30V polycrystalline
samples quenched from 800°C and aged as indicated.

400°C, or indirectly when the presence of ω-phase particles affects polycrystalline slip behavior sufficiently to induce the low energy fracture process within most grains as is shown in Fig. 7.

In summary, much of the influence of the microstructure on the fcp behavior of the age-hardenable Ti-V alloys examined can be viewed as being indirect; i.e., fcp can be adequately predicted by crack-tip opening models which simply take into account the changes of flow stress caused by the heat treatments. However, there are exceptions. Under certain conditions of precipitate morphology or distribution, such as when large Widmanstatten plates of the α-phase are present or when extensive α-phase occurs along grain boundaries, microstructure c affect fcp directly by providing a preferred path for locally rapid crack advance. The most pronounced (and least understood) effect of microstructure on fcp observed in this study occurs when ω-phase formation causes coarse, planar slip accompanied by cyclic softening. In single crystals, such slip affects the plane of cracking, while in polycrystalline specimens the slip behavior resulting from ω-phase formation is believed to be directly related to a "low energy" type of fracture across certain grains, and this in turn results in rapid overall fatigue crack growth.

The "low energy" fracture observed in the β Ti-V alloy fatigued in air suggests the possibility of an environmental effect with lab air when the ω-phase is present. An obvious possibility is some form of hydrogen-assisted cracking. In order to explore the effects of hydrogen on fatigue crack propagation of β Ti alloys, we are currently examining the fatigue crack growth behavior of a Ti-30V alloy both in the form of single crystals and polycrystalline specimens tested in dry helium and hydrogen gaseous environments.[17] Preliminary results indicate a change in fracture appearance occurs when cracking occurs in hydrogen gas. However, reliable crack growth measurements are not available at this time. Problems have arisen in constructing an environmental chamber capable of
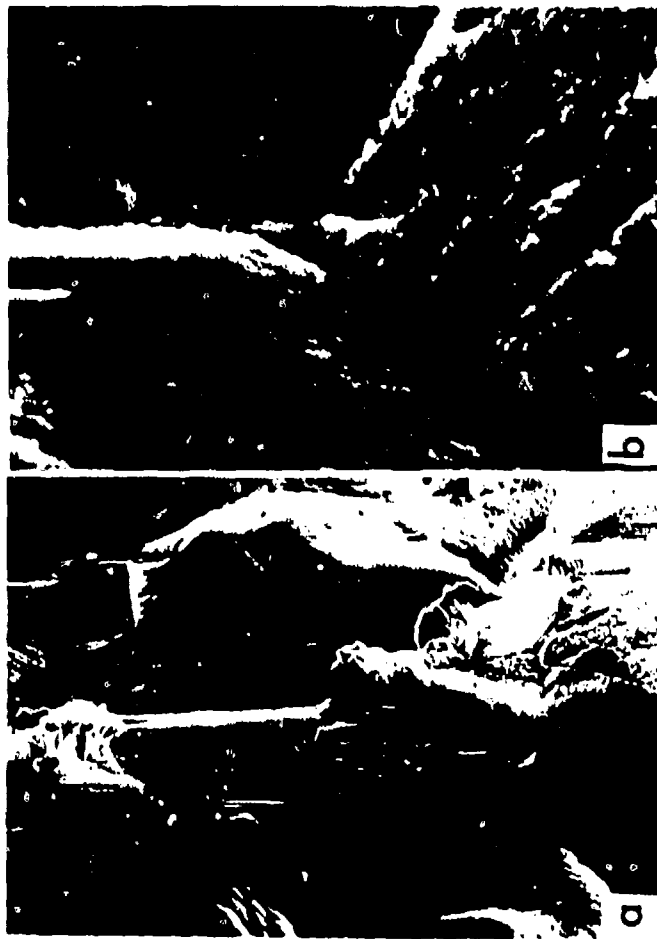
Figure 7. Electron micrographs showing the fracture surface of a Ti-30V polycrystalline sample heat treated at 300°C for 25 hrs. at: (a) 100X and (b) 1000X. $\Delta K \approx 11$ MPa$\sqrt{m}$.

sustaining a vacuum which is sufficient to remove residual gas effects prior to testing purified helium or hydrogen. This has retarded progress on this project but we anticipate completion this winter.

b. On the Cyclic Stress-Strain Response of Ti Alloys

In order to analyze the preceding fatigue crack propagation behavior in terms of the slip processes, the cyclic flow behavior of a Ti-V alloy was first characterized. Koss and Wojcik examined the cyclic stress-strain behavior of b.c.c. Ti-40 a/o V single and polycrystalline material.[18] In addition to obtaining values for the cyclic flow stress, which were necessary for a crack-tip plasticity analysis, they observed an asymmetry of the cyclic flow stress. For crystals with a stress axis near the 100 orientation versus those with a 110 stress axis, the flow stress (σ_{100} and σ_{110}, resp.) in tension (+) and compression (-) were:

$$(\sigma_{100})_- > (\sigma_{100})_+ \text{ and } (\sigma_{110})_+ > (\sigma_{110})_-. \tag{2}$$

These results are shown in terms of the cyclic stress-strain curves in Fig. 8. Such flow stress asymmetry is believed to be related to the mobility of screw dislocations which move more easily in the twinning direction than in the anti-twinning direction when sliding on a {211} plane.

Attempts have been made to relate cyclic work hardening behavior to fatigue crack propagation.[19,20] Such an approach appears reasonable in the case of low cycle fatigue behavior where the sample is fully plastic at plastic strains ∿1%. However, high cycle fatigue is characterized by a small plastic zone constrained by an elastic matrix and with rather intense slip occurring near the crack-tip where fresh surface for crack advance is being created. Thus there is poor correlation between the cyclic stress-strain response of Ti-40V[18] and the fatigue crack propagation behavior of the same alloy.[11] There is a recent study[21] that suggests cyclic stress-strain behavior measured at higher strain amplitudes may be meaningful insofar as crack propagation behavior.
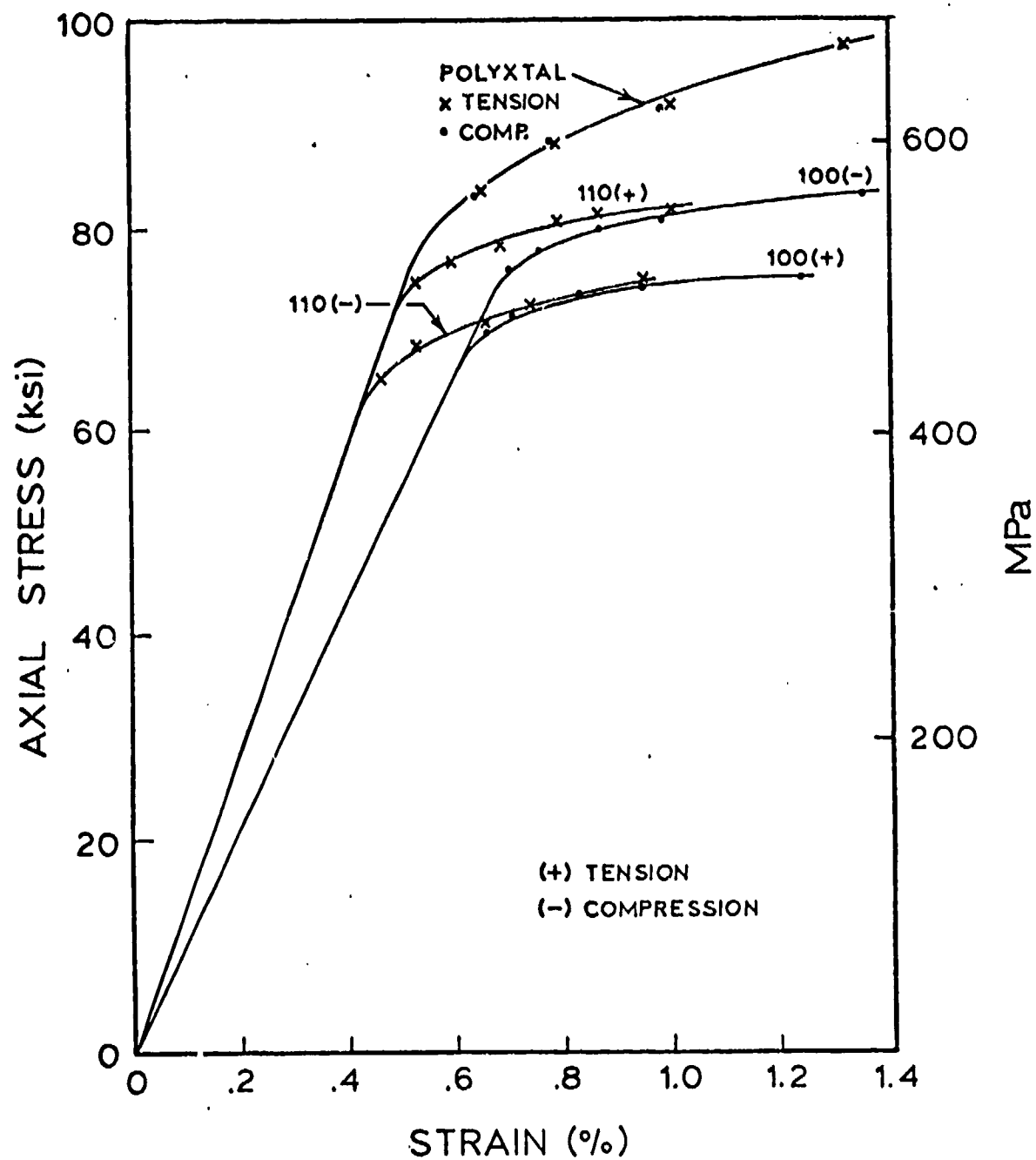
Figure 8. Cyclic stress-strain curves for Ti-40V alloy single crystals and polycrystalline material showing behavior in tension and compression.

In another attempt to relate cyclic stress-strain behavior to fatigue crack propagation behavior of Ti alloys, Wojcik and Koss briefly examined the cyclic stress-strain response of Ti-6Al-4V for three common microstructures: mill anneal, solution treat-and-age, and β anneal.[22] As shown in Fig. 9, all three microstructures exhibit at least an initial period of cyclic softening. Only the mill-annealed condition (MA) possesses well-defined cyclic stability with the peak stress saturating to a constant value after about 300 cycles. Thus one could obtain a valid cyclic stress-strain curve only for the MA condition. Since it is very common for high strength alloys not to show cyclic stability, this calls into question the usefulness of basing fatigue theories on cyclic stress-strain behavior, especially the cyclic work hardening rate, as determined from a valid cyclic stress-strain curve.

As before, relating the observed cyclic stress-strain behavior to fatigue crack propagation presents problems. Wojcik and Koss concluded that there was no simple correlation between the macroscopic cyclic stress-strain response and fatigue crack propagation in Ti-6Al-4V. However, there may be a correlation between cyclic stability and the fatigue ratio (as defined by fatigue strength at 10^7 cycles to ultimate tensile strength).

c. On the Origin of Microcracking Behind a Crack-tip

A common feature observed on the fracture surface of the β Ti alloys failed in fatigue is the presence of "secondary" or "slip-band" microcracks which intersect the main fracture surface; for example, see Figs. 3a and 10. Such microcracking is quite common during fatigue of many different alloys[11],[23-25] or even after stress corrosion cracking of an Al alloy.[26] Microcracking behind the crack-tip is usually observed on both faces of the fracture surface. However, Beachum[23] was the first to show that when the fracture surface is inclined to the stress axis, microcracking occurs only on the bottom face of an upward-sloping fracture surface. In the present work, Stout and Koss examined the distribution
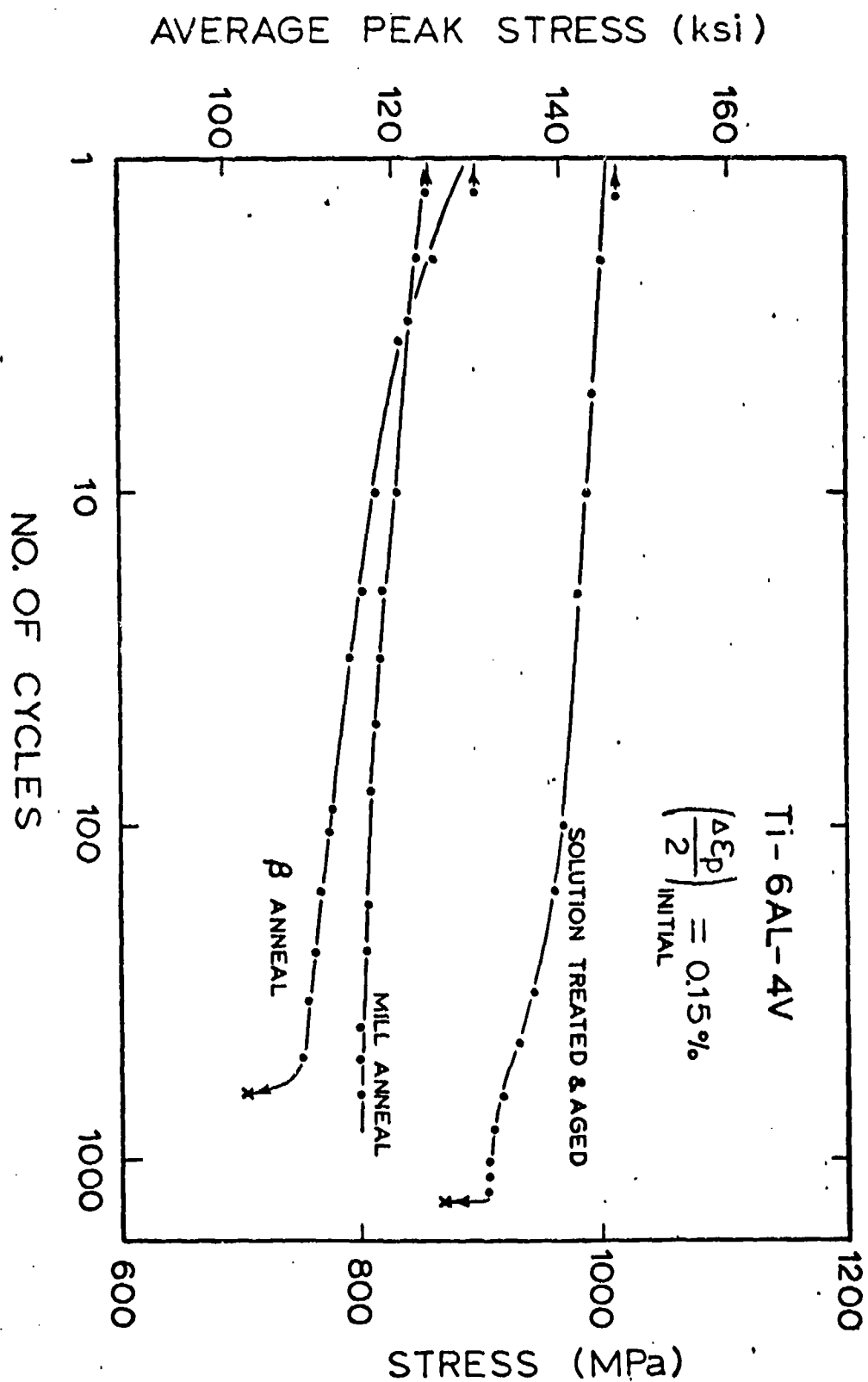
Figure 9. The cyclic stress-strain response behavior of the Ti-6Al-4V alloy.
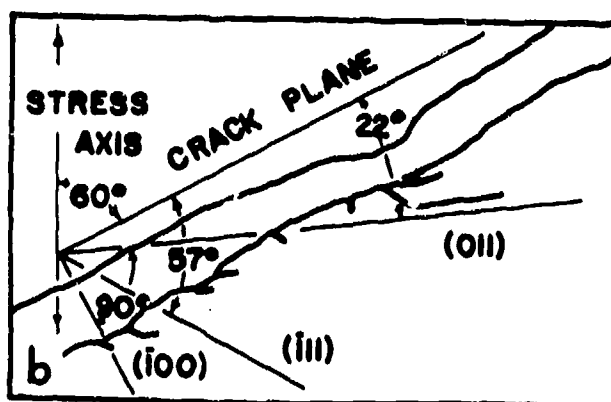
Figure 10. (a) Electron micrograph showing the profile of the main
fatigue crack shown in Fig. 3a. The schematic diagram
(b) depicts the crystallography of microcracking in (a).

of normal stresses around mode I as well as mixed mode I and II cracks in an elastic-plastic body with regard to the presence of microcracking behind a crack tip.[27]

Using Sih's analysis of small scale yielding in an elastic-plastic body with a crack,[28] we have calculated the radial distribution of the maximum normal stresses near a crack-tip. The results show that large tensile stresses exist behind the tip of the main crack on both faces for a mode I crack. However, large tensile stresses exist only on the bottom face of a mixed mode I and II crack as shown in Fig. 11. Since microcracking is observed only on the bottom face of an upward sloping crack, these calculations suggest that the observed microcracking is a result of the crack-tip stress field of the main crack.

In order to test more fully the concept that crack-tip stresses are the controlling factor in microcracking, the crystallography of microcracks which form during fatigue crack propagation of bcc Ti-V alloy single crystals was examined. In the samples analyzed, crack propagation occurred in a mixed mode I and II manner with the crack plane being inclined at an angle of ∿60° to the stress axis (as in Fig. 11). Microcracking was observed only on the bottom face of the upward sloping crack as in Figs. 3a and 10. An analysis of the crystallography of the microcracks shows that such cracking occurs on predominantly that {110} or {111} plane which calculations show has large tensile stresses normal to it. The magnitude of such stresses are also sufficient to cause local micro-fracture. Thus, Stout and Koss concluded that the state of stress induced by the main crack-tip provides a necessary and usually sufficient condition for the formation of microcracks behind a crack-tip.

That microcracking in the bcc Ti-40V alloys occurs on {110} and {111} planes with a large stress normal to them indicates that, at least in this alloy, the microcracking may represent cleavage on rarely observed planes. Such cleavage
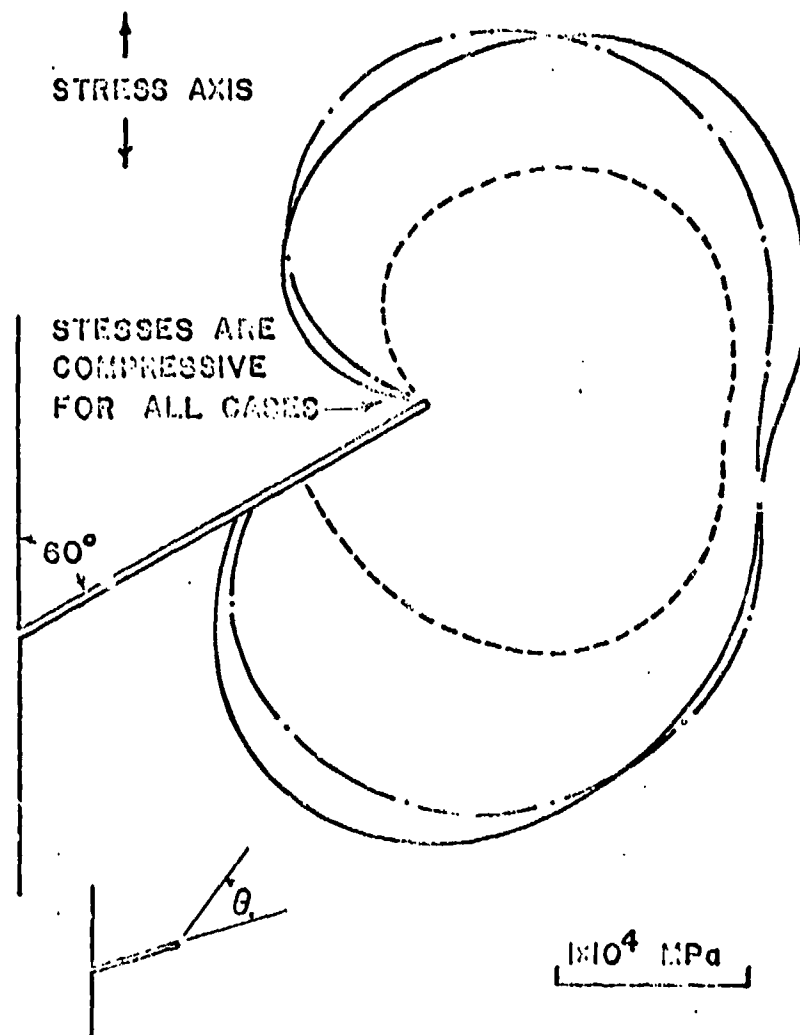
Figure 11. The radial distribution of the maximum normal stresses
calculated using crack-tip stress fields based on isotropic
and anisotropic elasticity as well as elastic-plastic
behavior. The conditions assumed are for a Ti-40V single
crystal in which : $a = 1.27 \times 10^{-3}$ m, $n = 33$, $\sigma = 300$ MPa,
$c = 848$ Mpa, and $r = 0.5$ μ.

may occur in special cases where large normal stresses are present and slip at the micro crack-tip is severely restricted by the stress field of the main crack.

d. Stage I Fatigue Crack Propagation in α and α-β Ti Alloys

While the fatigue behavior of the β Ti alloys appears in most respects very consistent with the behavior of fcc alloys and other bcc alloys, the fatigue crack propagation fcp behavior of hcp α or near-α Ti alloys is distinctly different. In these cases, if the scale of the microstructure is large compared to the crack-tip plastic zone, it is common to observe extensive Stage I crack propagation along a single crystallographic slip plane (rather than between the two active slip planes, which is the case in the β Ti alloys). Figure 12 shows an example of such crack propagation behavior across several large Widmanstatten colonies in a near-α, α+β Ti alloy. The previously described cyclic crack growth mechanism, which describes ductile fatigue in most bcc and fcc alloys, relies on a crack opening displacement normal to the crack surfaces. If, on the other hand, cracking occurs crystallographically along a slip plane and slip at the crack tip is predominantly confined to that single system (as appears to be the case for Stage I cracking in α-phase Ti alloys), then there is very little crack opening displacement (COD) normal to the crack plane; nearly all of the COD is the form of sliding of the crack faces. This strongly suggests that the mechanism of crystallographic Stage I cracking in the α-phase Ti alloys (as well as Ni base superalloys) differs from that of ductile fatigue in β Ti and other bcc and fcc alloys. In the former, fatigue appears to be a result of deformation confined primarily to a single slip system, while in the latter at least two slip systems are required.

Utilizing specimens consisting of individual Widmanstatten colonies of the Ti-8Al-1Mo-1V alloy, for the first time ever the effect of crack orientation on crystallographic Stage I fatigue crack propagation has been studied.[29] Crack propagation is characterized by cracking almost exclusively on the basal plane
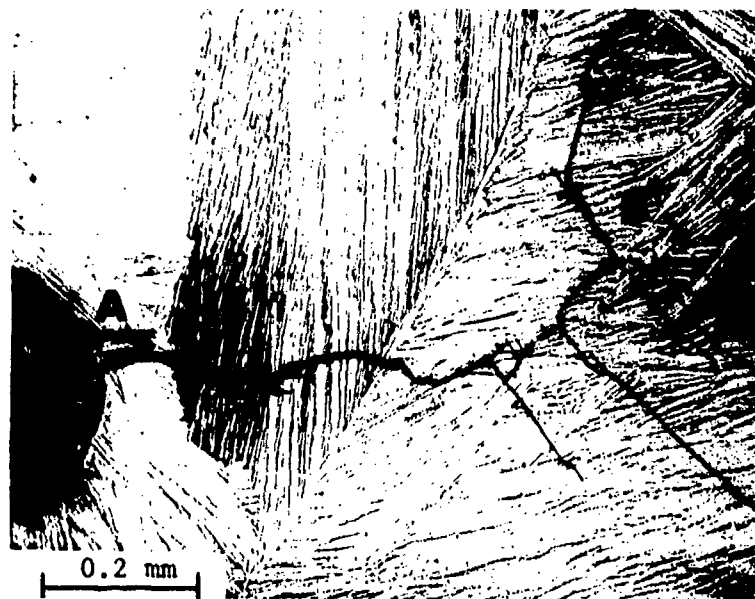
Fig. 12. Fatigue crack initiation (A) and propagation (B) in a
center-notched IMI 685 fatigue crack growth specimen.
After Shechtman and Eylon (ref. 34).

of the hcp α-phase in this α+β alloy. As such, the orientation of the Stage I crack plane to the stress axis is dependent on colony orientation. The dependence of the crack propagation rates on colony (and therefore crack) orientation is shown in Fig. 13. As can be seen, crack growth is most rapid when the crack plane is inclined to the stress axis with the propagation rates in lab air being a strong function of crack orientation. Crack propagation in a dry helium gas environment is characterized by crack growth rates which are five to ten times slower than in lab air (see Fig. 14) and which are not as sensitive to crack orientation as in lab air.

Fatigue cracking in air is characterized by cleavage-like fracture surfaces which river markings suggesting that crack growth occurs in the <$11\bar{2}0$> directions (see Fig. 15). On the other hand, fatigue crack propagation in a helium environment results in fracture surfaces which show considerable ductility (compare Fig. 15b to Fig. 16). These results clearly indicate that this form of Stage I fatigue crack propagation is quite sensitive to environment, probably to the water vapor in the lab air. We have also analyzed the plasticity ahead of the crack-tip in terms of the mixed mode cracking and single crystal plasticity. While this method works well for analyzing the fatigue behavior of Ti-V alloy single crystals, at this time we are unable to establish a clear-cut correlation between predicted behavior based on our calculations and the observed crack growth rates. We are performing further analysis in the hopes of establishing a mechanism for such fatigue behavior.

Although small amounts of β phase dispersed within the α-phase in the form of lamellar is known to influence the deformation behavior greatly,[30] the previous results suggest that Stage I cracking is dominated by the behavior of the α-phase. In a study parallel to that described above, Stage I fatigue crack propagation has been examined in α-phase Ti-7 a/o Al alloy single crystals.[31] Testing specimens of similar crystallographic orientation in lab air as well as
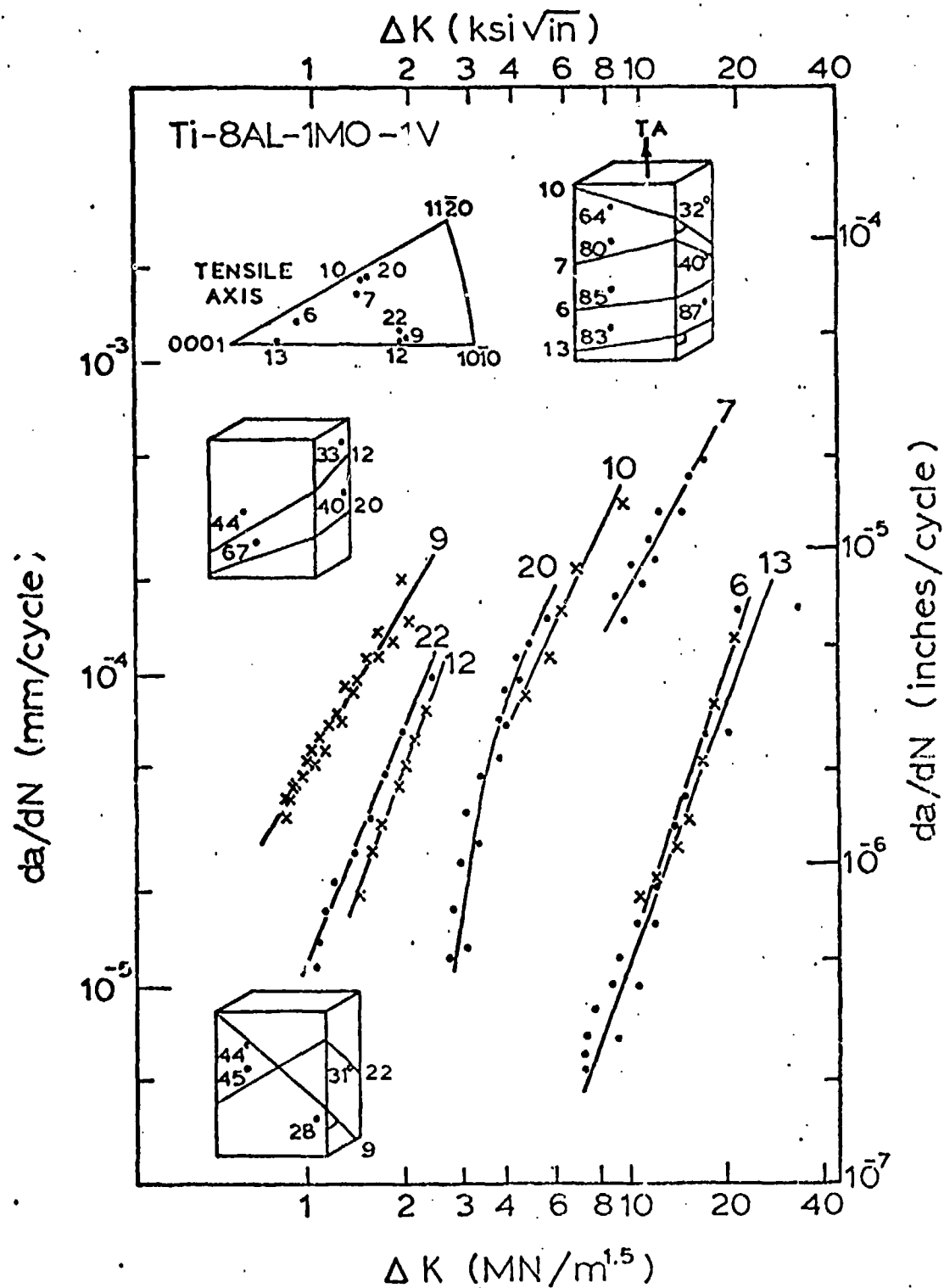
Figure 13. The dependence of the fatigue crack growth rate on the stress intensity range for individual colonies of a Ti-8Al-1Mo-1V alloy tested in lab air, R. H. ≈ 20% The orientations of the stress axis and of the fracture planes are also shown. Samples 10, 12, and 13 were solution-treated at 925°C, and all of the others were solution treated at 760°C.
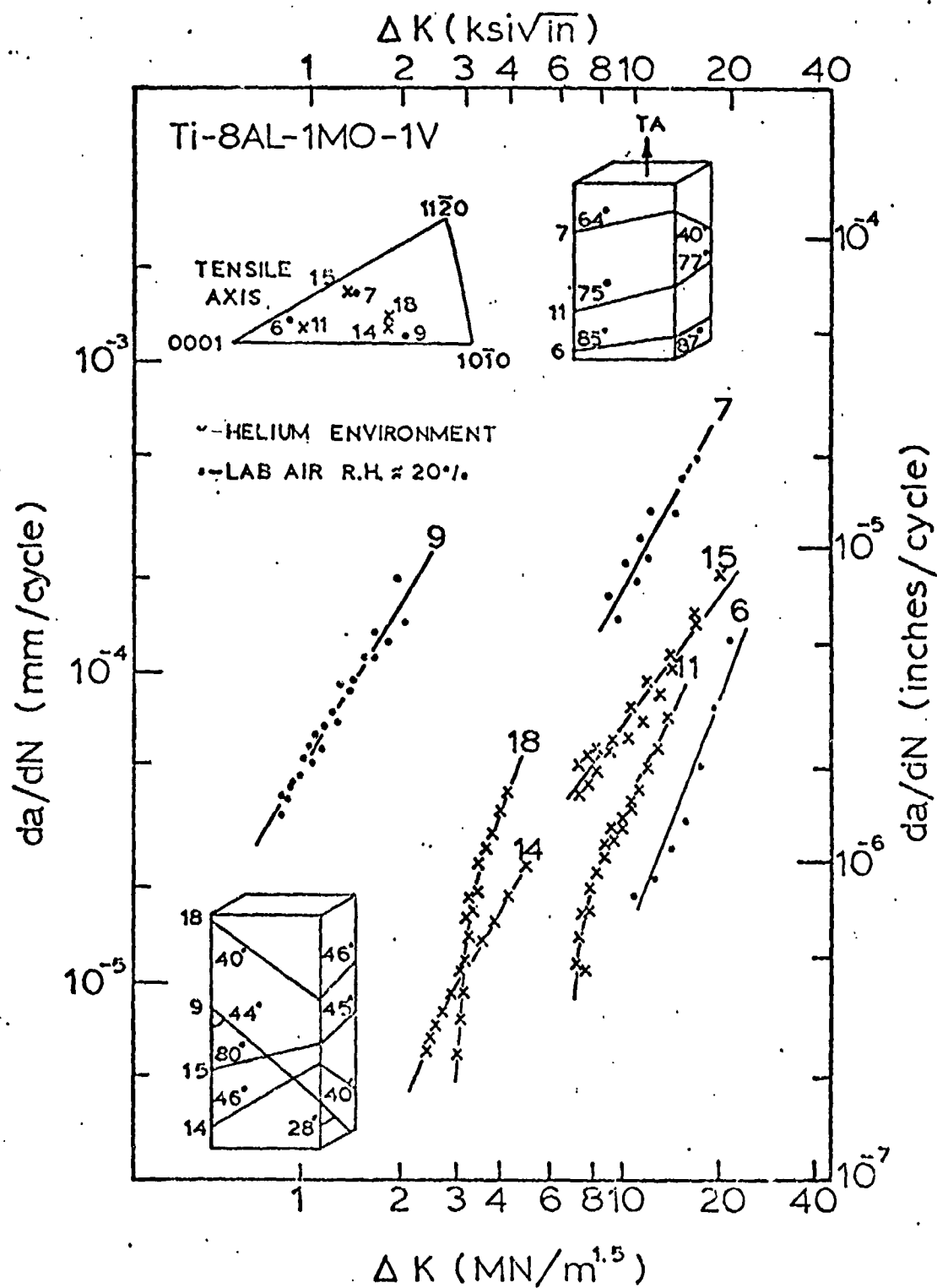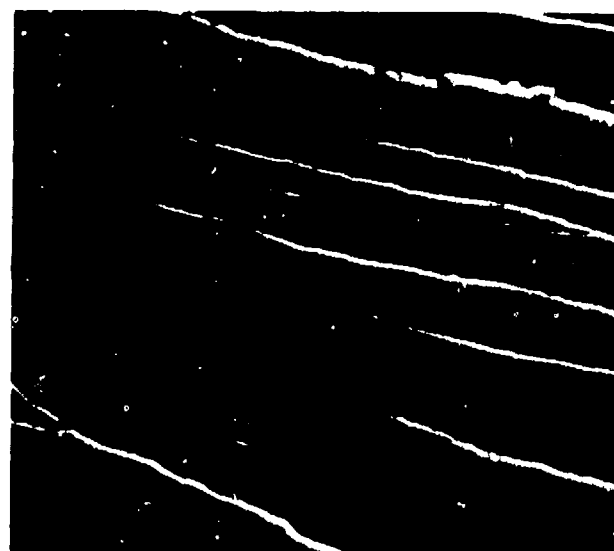
Figure 14. The dependence of the fatigue crack growth on stress intensity range for individual colonies of Ti-8Al-1Mo-1V colony specimens tested in dry helium and lab air environments. The orientations of the stress axis and fracture planes are shown. All samples were solution treated at 760°C.

a) Composite fractograph
   20X

b) $\Delta K = 2.2$ MPa$\sqrt{m}$ 500X

   da/dN $= 2 \times 10^{-6}$ mm/cycle

$<11\bar{2}0>$

CPD

Figure 15. Scanning electron fractographs of speci...n No. 9 (see Fig. 14) tested in lab air.

Figure 16. Scanning electron fractograph of specimen No. 11 (see Fig. 14) tested in dry helium at ΔK = 11 MPa$\sqrt{m}$. The crack propagation direction (CPD) and the orientation of the β plates are indicated. 500X.

in dry helium shows that both crystallographic orientation and environment affect crack growth rates. This is shown in Fig. 17 where in all cases shown the crack plane is the (0001) plane. Confirming the behavior of the Ti-8Al-1Mo-1V colonies, the Ti-7Al specimen shows the slowest rates at Stage I cracking when the crack plane tends toward being perpendicular to the stress axis.

Fig. 17 also shows that lab air is an aggressive environment insofar as Stage I cracking in this α-phase alloy. For a given specimen and therefore crack plane orientation, the results in Fig. 17 show that dry helium increases the threshold stress intensity necessary for slow crack growth and decreases the fast crack growth rates by a factor of about seven. As with the Ti-811 behavior, a transition in fractography occurs from cleavage-like in air to a more ductile fracture appearance for samples failed in helium. We are also analyzing this Ti-7Al fatigue data in terms of crack-tip plasticity with the goal of establishing the mechanism of crystallographic Stage I fatigue crack propagation.

That the lab air environment should accelerate Stage I fatigue crack propagation in α and α-β Ti alloys suggests hydrogen, resulting from the presence of water vapor, may be an important factor in such cracking. We have therefore initiated a study of the influence of gaseous hydrogen on the fatigue crack propagation of Ti-7Al single crystal and polycrystalline specimens.[32] As with the study of fatigue of a Ti-V alloy in hydrogen, this project has been slowed by experimental problems in construction of an environmental chamber. We anticipate completion of this study late this winter.

e. Hardening Ti Alloys by Eutectoid Decomposition

In a small separate project, the hardening of Ti alloys by eutectoid decomposition was briefly surveyed.[33] The formation of bainite by eutectoid decomposition is a well-known method of hardening steels. Similar reactions occur in a wide number of Ti alloys and the accompanying phase transformations are being investigated by Prof. H. Aaronson and co-workers at Michigan Tech. In an effort to
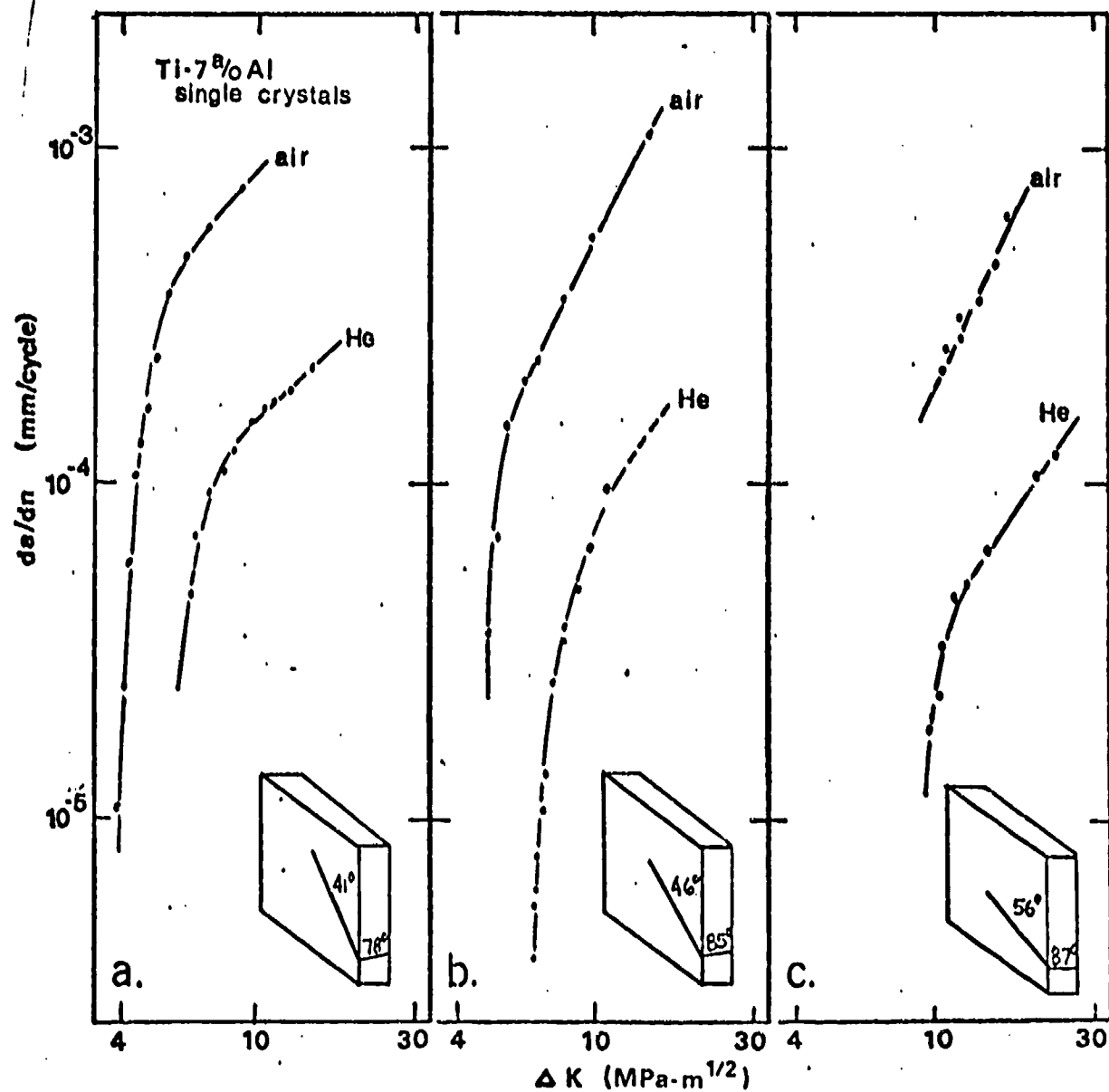
Figure 17. The dependence of the fatigue crack growth rate on the stress intensity range for single crystals of the α-phase alloy Ti-7 a/o Al. tested in lab air (R.H. ≈ 20%)and a helium and at R = 0.1. The orientation of the crack plane to the stress axis is also shown.

examine the possibility of obtaining high strengths in Ti alloys utilizing a bainitic microstructure, we have surveyed the age hardening response of selected Ti-Co, Ti-Ni and Ti-Cu as well as Ti-V-Ni and Ti-V-Cu alloys which have been directly quenched from 1000°C into lead at 500°-700°C and isothermally reacted to form bainite. The binary alloys are soft (DPH $\leq$ 225 kg/mm^2) and cannot be hardened appreciably because the bainite which forms is very coarse due to the high aging temperature necessary to avoid martensite formation. In order to refine the scale of the microstructure, we briefly examined ternary alloys in which V was added to suppress the martensite start M_s temperature to a temperature ($\sim$500°C) low enough to permit more favorable conditions for the nucleation and growth of a fine bainitic microstructure. The age hardening response of these (Ti-V)-X alloys at 500°C is shown in Fig. 18 and indicates that a modest increment in hardness (Δ DPH $\simeq$ 40 to 50) occurs upon reacting for short times ($\sim$2 hours) compared with that of the coarsened microstructure at long aging times. Higher hardnesses and a larger hardening response should be possible in alloys with a higher V content but an even lower aging temperature would be necessary.

DPH

400

300

200

AGING TIME (hrs)

2   4   10   20   40

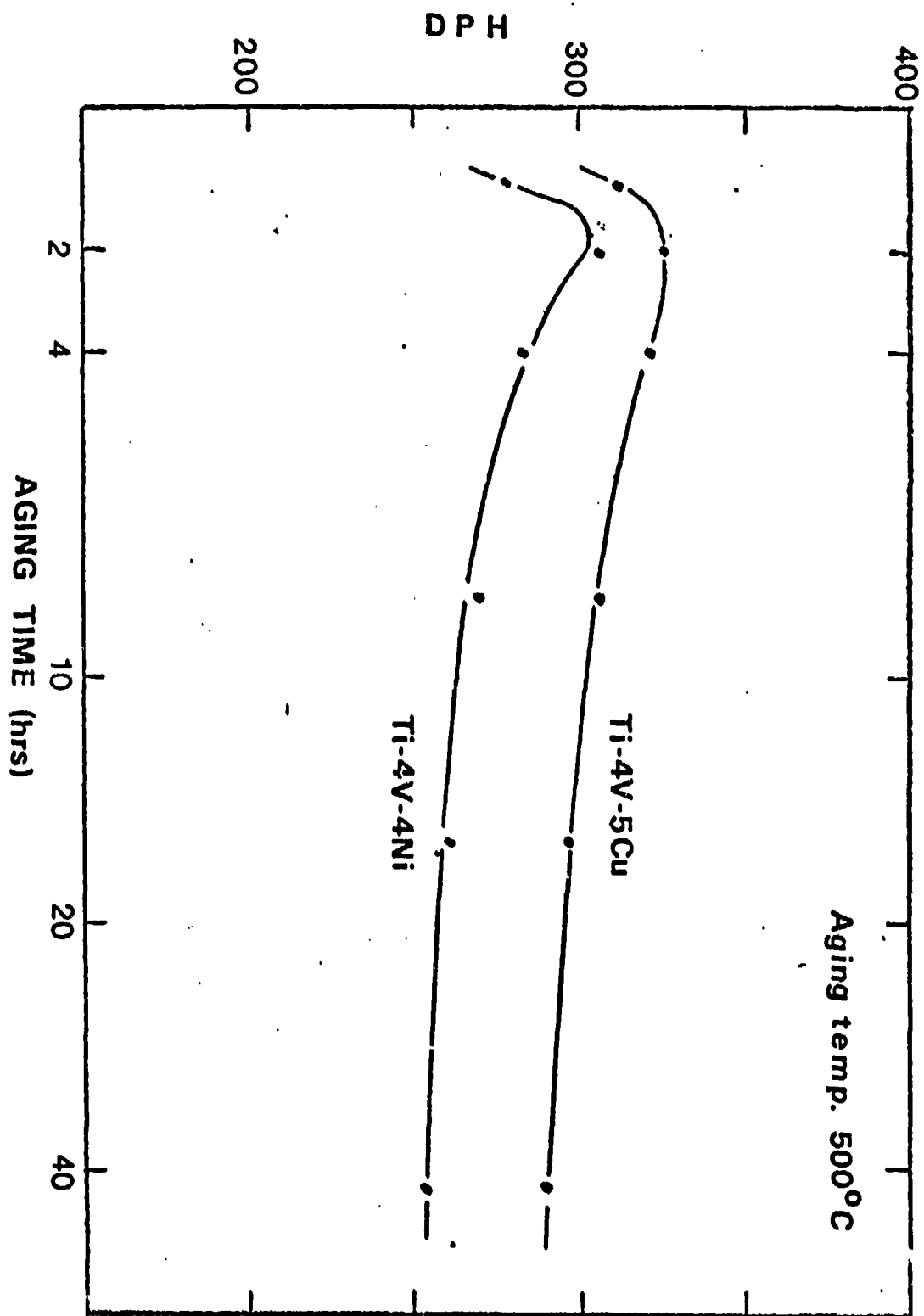Aging temp. 500°C

Ti-4V-5Cu

Ti-4V-4Ni

Figure 18. The hardness response upon eutectoid decomposition of two alloys solution annealed at 1000°C and directly quenched and reacted at 500°C. The alloy compositions are in wt %.

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"The Influence of Gaseous Hydrogen on the Fatigue Crack Propagation Behavior of
   a Metastable Ti-V Alloy," (with J. Wilcox), work in progress; to be submitted
   for publication in 1979.

"Crack Propagation Behavior of Ti-Al Single Crystals in Gaseous Hydrogen,"
   (with L. Petzold), work in progress; to be submitted for publication in 1979.

Dissertations

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